

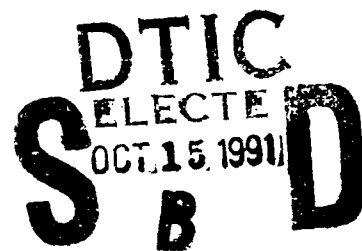
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PRECIPITATION HARDENING OF A BETA TITANIUM
ALLOY BY THE ALPHA-TWO PHASE

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PRECIPITATION HARDENING OF A BETA TITANIUM ALLOY BY THE ALPHA-TWO PHASE

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Abstract

The age-hardening of beta titanium alloys by the formation of ordered alpha-two precipitates based on Ti_3Al has been investigated by transmission electron microscopy and hardness observations. Results of tests based on the alloy Ti-23Nb-11 Al (at.%) show a large precipitation hardening response at temperatures considerably higher than is possible in current beta titanium alloys. TEM identifies the hardening to be caused by the formation of ordered, alpha-two precipitates with a lath-like morphology.

Introduction

As a class, most metastable or near-beta Ti alloys are of commercial significance because of an age-hardening capability which depends on the formation of hcp alpha-phase precipitates [see ref. 1, for a review] during the aging of solution-treated material in the 450°-500°C range. Despite continued interest in commercial applications of beta titanium alloys, most recent developmental efforts have focused on titanium alloys based on either the alpha-two Ti_3Al or the gamma $TiAl$ phase. The present communication reports initial results which indicate that it is possible to age-harden the beta Ti alloy, Ti-23Nb-11Al (atomic %), with alpha-two precipitates.

The age-hardening response, which is quite pronounced, is such that peak hardness occurs at $\sim 575^{\circ}\text{C}$ or about 100°C higher than that observed in conventional, age-hardenable beta Ti alloys. The purpose of this communication is to identify pertinent characteristics of this new age-hardenable alloy system.

Experimental

The alloy examined had a composition of Ti-22.8Nb-11.1Al in at. % or Ti-35Nb-5Al in wt.%; hereafter it will be referred to as Ti-23-11. The specimens, which contained 1580 wt ppm oxygen, originated from bar stock which was extruded at 1038°C , using a 22:1 extrusion ratio. All heat treatments were performed by encapsulating the specimens in quartz under a partial pressure of high purity argon with tantalum foil being wrapped around the specimens to getter any residual oxygen. Transmission electron microscopy (TEM) examinations were performed on foils thinned by ion milling using a gun angle of 15° for 10 hours, followed by 2 hours of milling at 5° . Ion milling was performed using a voltage of 3.5 kV and a current of 1mA.

Results and Discussion

(a) Solution-Treated Condition

In order to determine a solution-treatment temperature for age hardening, specimens which had previously been slow cooled from 1100°C were heated to temperatures ranging from 930° to 1035°C in 15°C increments and quenched into an ice-water bath. Care was taken to break the quartz capsule containing the specimen immediately after quenching. The results indicated that heat treatments of 960°C or higher resulted in a single phase microstructure while at 945°C coarse precipitates were retained. Thus a solution-treatment temperature of 1000°C followed by an ice-water quench was used. The above solution-treatment resulted in a grain size of $210\text{ }\mu\text{m}$.

The following observations were made from TEM of solution-treated specimens. First, the absence of superlattice reflections that are characteristic of the ordered beta phase with a B2 structure indicates that the Ti-23-11 alloy is disordered in the solution-treated condition. While an ordered B2 phase has been observed in other Ti-Nb-Al alloys (2-5), it should be noted that these

contain more Nb and Al than the present alloy. Secondly, the omega phase was not observed. This is probably a result of the significant amount of Al (11a/o) present; previous research has shown Al is effective in suppressing omega-phase formation (6). Thirdly, diffuse streaking is apparent in the $\langle 110 \rangle$ β directions of the $\langle 100 \rangle$ β zone axis selected area diffraction SAD pattern. Such streaking has been observed in other Ti-Nb-Al alloys (2-4) and has been attributed to a $\{110\} \langle 100 \rangle$ lattice instability in bcc-type lattices (7). Thus, we conclude that the Ti-23-11 alloy retains a disordered bcc lattice upon quenching from 1000°C; athermal omega phase is not present.

(b) Age Hardening Response

As shown in Fig. 1, aging the solution-treated Ti-23-11 alloy in the 575°-625°C range results in a significant increase in hardness from ~270 VHN to ~440-470 VHN. The age hardening occurs rapidly with most of the hardness increase occurring after 3 hours at temperature. In contrast, overaging after peak hardness is attained occurs rather slowly, especially at 575°C, which suggests rapid nucleation and growth of the hardening phase but sluggish particle coarsening kinetics.

While the age-hardening response over a wide range of temperatures has not yet been determined currently for the Ti-23-11 alloy, results of tests conducted on material aged 6 hours at temperatures from 375° to 675°C indicate that peak hardness occurs at 575°C. As such, this is about 100°C higher than is typical of conventional beta Ti alloys (1).

(c) Microstructural Changes and Age Hardening

Preliminary TEM was confined to the condition in which the specimen was solution treated and subsequently aged at 575°C for 6 hrs. For this condition, Fig. 2 shows that in the bright field imaging mode, elongated precipitates are present with a tendency to form in a zig-zag morphology. This type of morphology is similar to that previously observed in an investigation of orthorhombic precipitates in an ordered bcc matrix of an alloy based on Ti_2AlNb ; in that study, the zig-zag morphology was interpreted to be a result of minimization of elastic energy (8).

An identification of the precipitate structure is based on the indexing of selected area diffraction SAD patterns such as those shown in Figs. 3 and 4. The presence of precipitates in a

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bcc matrix is indicated in the $\langle 111 \rangle \beta$ zone axis SADP shown in Fig. 3 and the $\langle 110 \rangle \beta$ zone axis SAD pattern in Fig. 4. As expected, an indexing of the more intense matrix spots reveals they are caused by a disordered bcc-type lattice. An analysis of the precipitate reflections in the $\langle 111 \rangle \beta$ SAD pattern of Fig. 3 indicates that indexing of precipitate reflections is consistent with an hcp structure; this is depicted in the computer-simulated diffraction pattern of the $\langle 111 \rangle \beta$ zone axis also shown in Fig. 3. However, the possibility exists that the lattice may have an orthorhombic distortion and might be based on Ti_2AlNb . In order to properly identify whether or not the precipitates are alpha-two phase, a $\langle 110 \rangle \beta$ zone axis SAD pattern was indexed (9). The $\langle 110 \rangle \beta$ zone axis SAD pattern in Fig. 4 exhibited diffuse streaking or spiking at the precipitate reflections, which makes the precipitate lattice difficult to index. However, since it is already known that the precipitates have an hcp or orthorhombic-distorted hcp structure, only the inner ring of precipitate reflections next to the transmitted beam reflection are needed for structural determination. Figure 4 indicates that measurement of the angles between precipitate reflections of a single variant are 60° , which indicates that the precipitates are the alpha-two phase. If the precipitates had an orthorhombic structure, the corresponding angles would be 58° and 63° . Since the angles are $60 \pm 1^\circ$, we conclude that the precipitates are alpha-two phase based on Ti_3Al , which has a DO_{19} structure.

In addition, the $\langle 110 \rangle \beta$ zone axis (in Fig. 4) shows that, similar to 'Burgers'-alpha phase in other beta Ti alloys (10), the alpha-two particles are crystallographically related to the bcc β matrix with the Burgers orientation relationship:

$$\begin{aligned} \{0001\} \alpha_2 // \{110\} \beta \quad \text{and} \\ \langle 2\bar{1}10 \rangle \alpha_2 // \langle 1\bar{1}1 \rangle \beta . \end{aligned}$$

It might be noted that, given the above orientation relationship, twelve orientational variants of the alpha-two precipitates are present in the matrix, contributing to the complexity of the SAD patterns.

Figure 5a is a centered dark-field (CDF) image showing two variants of the alpha-two phase precipitates. The precipitates appear as laths viewed nearly edge-on in this orientation. In Figure 5b, a CDF image, obtained under different tilt conditions, reveals the lath-like nature of the

alpha-two phase precipitates. In this view the precipitate laths are seen face-on. By measuring the dimensions of the precipitates from CDF images under different tilt conditions, the average dimensions for the precipitates are given as $a \simeq 170$ nm, $b \simeq 120$ nm, and $c \simeq 30$ nm. Therefore, the precipitate morphology can best be described as lath-like since $a > b \gg c$.

Summary

A Ti-23Nb-11Al alloy with a disordered beta matrix has been age-hardened by the formation of lath-like, alpha-two precipitates based on Ti₃Al. The age-hardening, which is quite pronounced, occurs at a temperature range of 575-625°C, which is significantly higher than that which occurs in conventional beta Ti alloys age-hardened by alpha-phase precipitates.

Acknowledgments

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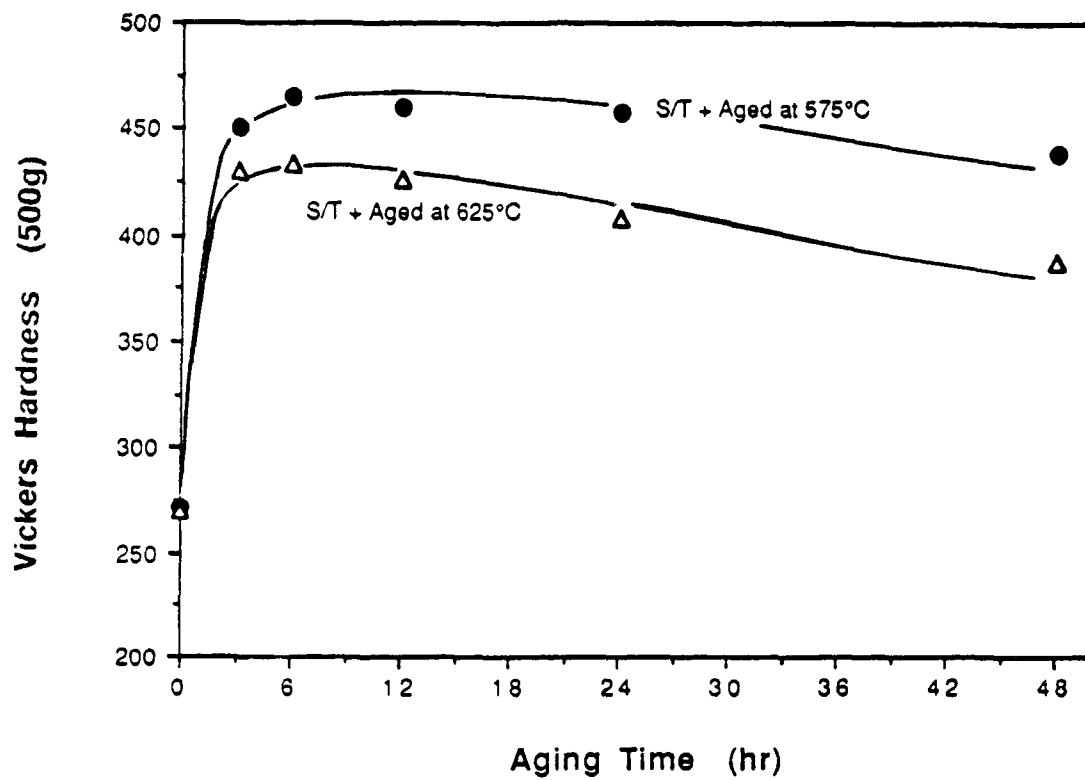


Figure 1. The age-hardening response of Ti-23Nb-11Al solution-treated (S/T) at 1000°C, quenched, and aged at either 575° or 625°C.

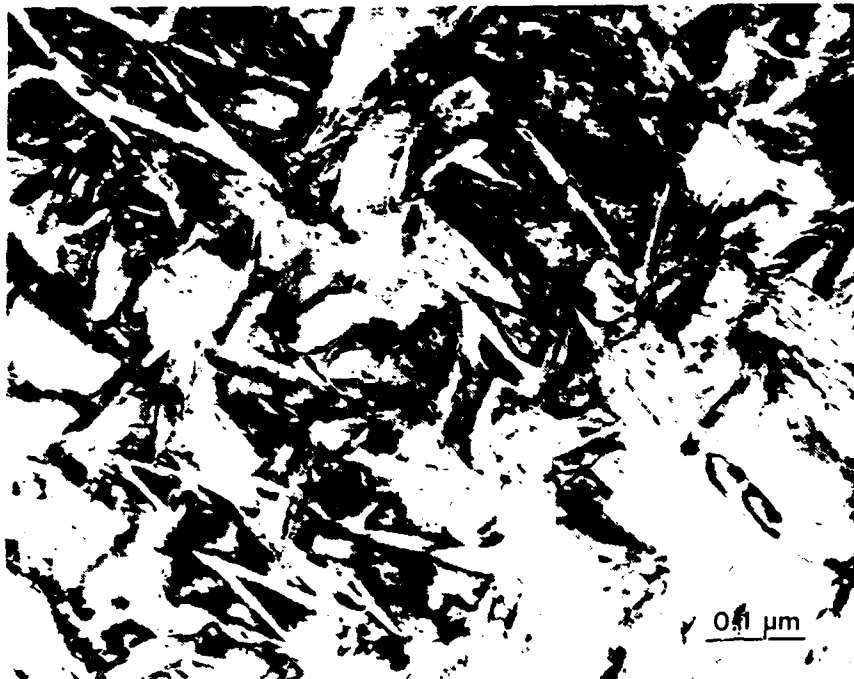


Figure 2. A TEM bright field image of the alloy aged at 575°C for 6 hours, showing precipitates arranged in a zig-zag fashion.

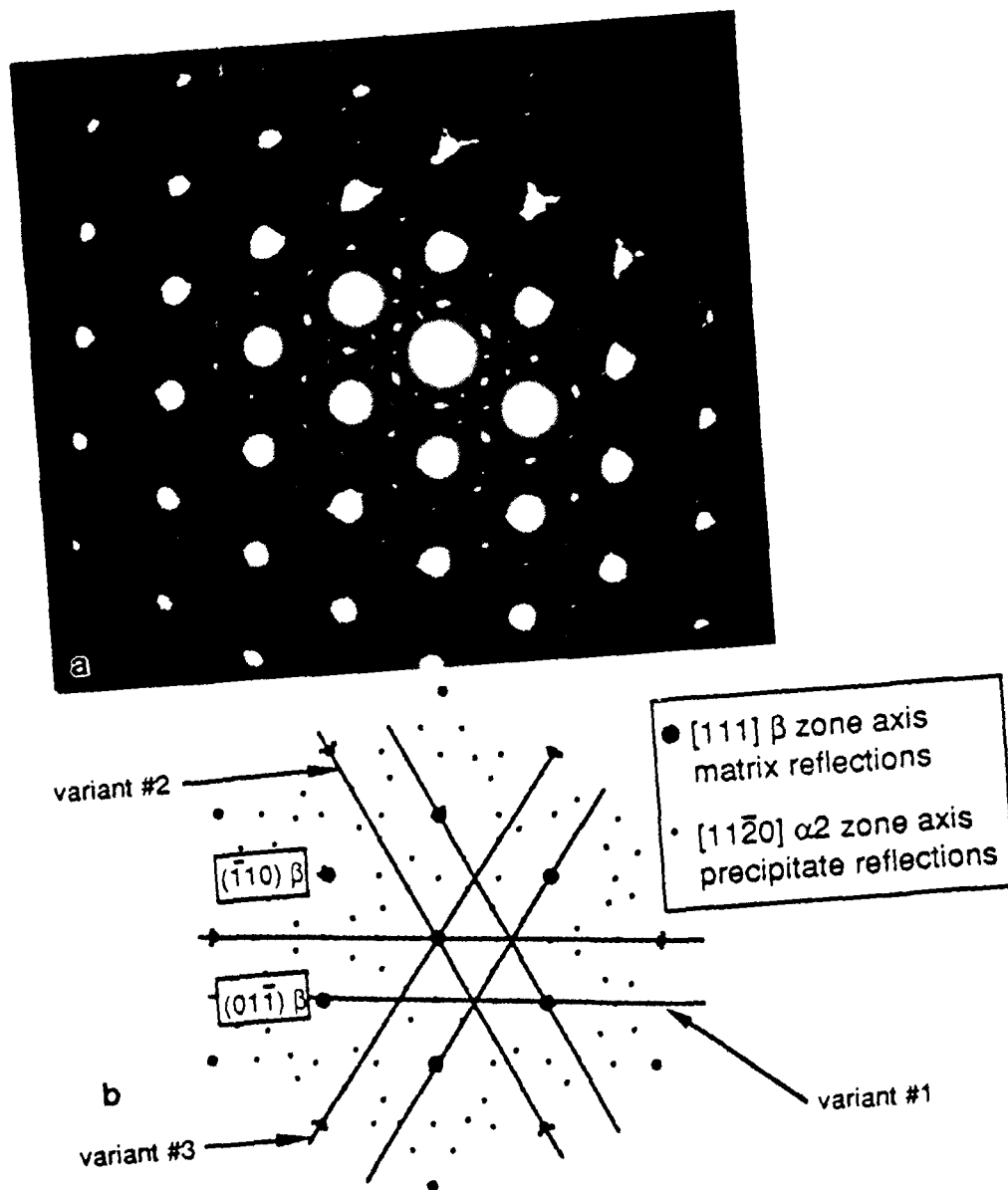


Figure 3. SAD pattern of the Ti-23-11 alloy aged at 575°C for 6 hours; (a) $\langle 111 \rangle \beta$ zone axis and (b) computer-simulated $\langle 111 \rangle \beta$ zone axis.

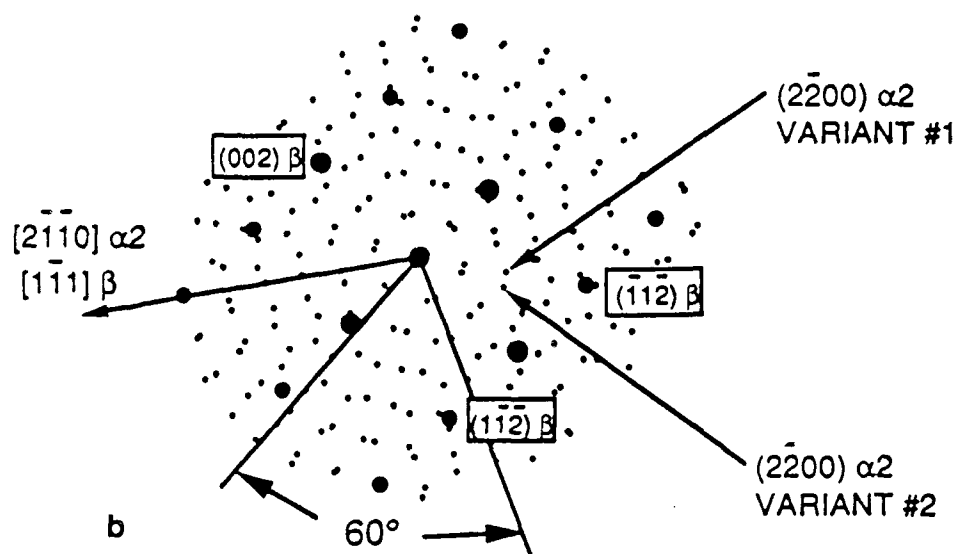
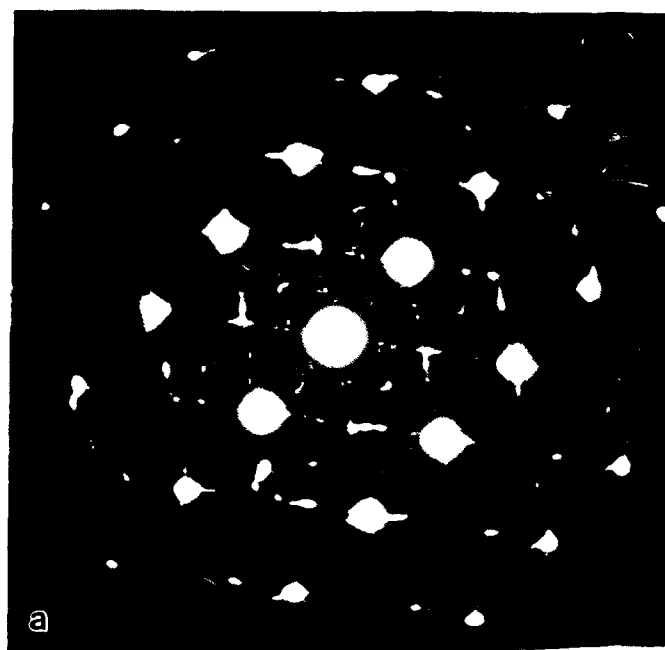


Figure 4. SAD pattern of the Ti-23-11 alloy aged at 575°C for 6 hours; (a) $\langle 110 \rangle \beta$ zone axis and (b) computer-simulated $\langle 110 \rangle \beta$ zone axis.

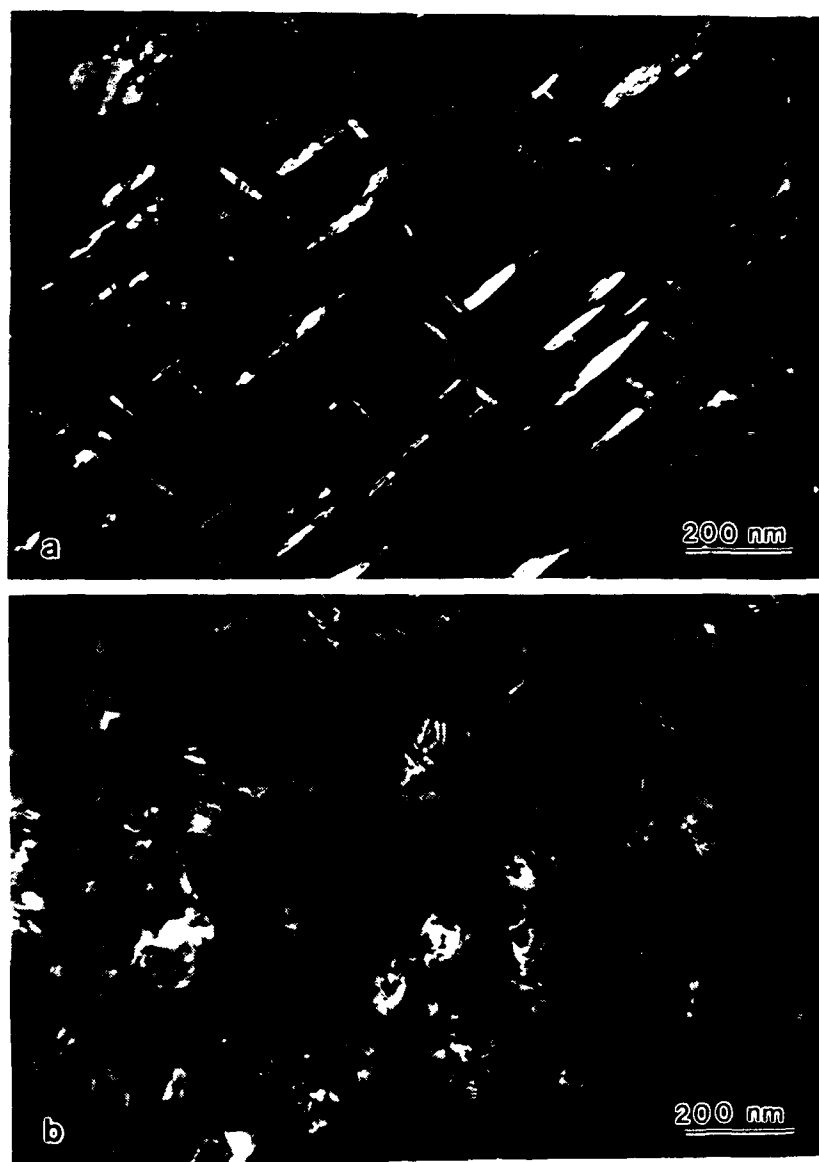


Figure 5. Centered dark-field image of the alpha-two precipitates. Two variants are seen in (a), while a single variant is shown in (b). The alpha-two precipitates formed on aging the alloy at 575°C for 6 hours.